CRACKING MECHANISMS IN HY-130 STEEL WELDMENTS.

Alan Robert Becker

CRACKING MECHANISMS IN HY-130 STEEL WELDMENTS

by

ALAN ROBERT BECKER

B.S., United States Naval Academy

(1965)

Submitted in partial fulfillment of the requirements

for the degree of

OCEAN ENGINEER

at the

Massachusetts Institute of Technology

June 1973

ABSTRACT

CRACKING MECHANISMS IN HY-130 STEEL WELDMENTS

by

ALAN ROBERT BECKER

Submitted to the Department of Ocean Engineering on May 11, 1973, in partial fulfillment of the requirements for the Degree of Ocean Engineer.

The objective of this research was to study the causes and mechanisms of cracking in HY-130 steel weldments.

The various modes of crack initiation were investigated by use of a high temperature tensile test with a GLEEBLE device. The fractured surfaces were examined metallographically by use of the scanning election microscope. The results were then compared to cracks found in the metallographic examination of a circular fillet welded specimen.

Cracking was found to be a three-stage phenomena. Intergranular cracks form in the heat affected zone of the weld at or near the fusion line where the temperature is just below the melting temperature. These intergranular cracks intersect with sulfide inclusions which appear at the grain boundaries as the temperature decreased and the grain distortion increased. The sulfide inclusion appears to act as a plastic hinge causing further cracking to extend transgranularly.

The actual welding cycle employed appears to have little influence on the severity of the cracking process.

A proposal is offered to reduce the probability of cracking by reducing the size of the sulfide inclusions.

Thesis Supervisor: Koichi Masubuchi
Title: Professor of Ocean Engineering

ACKNOWLEDGMENTS

I wish to thank the National Science Foundation and the Alcoa Foundation who jointly sponsored this thesis effort. Several organizations and individuals were of particular assistance. Mr. Abraham Pollack of the Naval Ship Research and Development Center, Ferrous Welding Laboratory, Annapolis, provided the materials and Mr. John Langone of the Boston Naval Shipyard provided all the necessary machining services.

The Army Materials and Mechanics Research Center, Watertown, was kind enough to permit the use of the Gleeble tester for the experimental testing and Mssrs. Don Buffum and Joe Ianelle of that organization were especially helpful in running the test program.

At M.I.T., Mr. Arthur Gregor provided much needed advice as well as assistance in the metallographic preparation for this work. Professor Koichi Masubushi has been my advisor for some three years now, for which I am certainly indebted. I wish to also thank Professor Arden Bement whose discussions and advice provided the insight necessary for this thesis.

Finally, I thank Ms. Louise Carella for her efficient typing of this thesis.

TABLE OF CONTENTS

LIST OF TABLES

LIST OF FIGURES

LIST OF FIGURES

INTRODUCTION

Although the introduction of higher yield strength steels has broadened the working limits of welded structures such as pressure hulls of submersibles and high pressure vessels, it has also broadened the problems in the design and fabrication of these structures. With the increased yield stress level, there is a corresponding decrease in fracture toughness, thus emphasizing the need for flaw free welded joints. As it is becoming increasingly evident that it is virtually impossible to produce a flaw free welded structure, the problem of weld cracks increases in importance, especially in view of the relatively low notch toughness exhibited by the high yield strength steels.

Beginning with the introduction of HY 80 into widescale shipbuilding use in the mid 1960's, the problem of weld cracking became the center of considerable research. HY 80 was created as a low alloy steel with a minimum required yield strength of 80,000 psi. Its strength and toughness were derived from quenching and tempering as compared with its predecessor (HTS) which derived its strength from cold rolling.

In the early stages of use of HY 80, it became clear that existing welding procedures and testing were insufficient to insure a sound, reliable joint. From the HY 80 problem arose a new philosophy of material testing which

will lead to the ultimate acceptance or rejection of the proposed new steel.

Basically, the philosophy calls for the initial certification of the material to meet the requirements of strength, toughness and composition. Passing this, the next stage is weldability testing where welding procedures and welding materials are certified. The final requirement is for "semi work scale" testing where the material is tested in welded sections of full scale thickness under conditions simulating multiaxial loads. It is important to note that each stage must be successfully completed prior to any further development.

HY 130 steel is the new generation high strength steel presently under consideration and testing for future use. As was the case with HY 80, HY 130 was expected to experience and is exhibiting weld cracking problems. However with the experience gained in the HY 80 program, this problem can be attacked and will hopefully be solved prior to the introduction of this steel into widescale use.

This work attempts to characterize weld cracking in HY 130 to determine the basic causes and mechanisms of cracking in HY 130 weldments.

From the results of several studies made of the cracking mechanisms in welded structures,(1,2) it is thought that microcracks are created at high temperatures and grow

to form macrocracks under conditions of high restraint at low temperature. These conditions of high temperature and high restraint occur in the weld metal or in the heat affected zone (HAZ). Generally speaking, cracking of the unaffected base metal has not been found to be a problem.[3] Additionally, in the work of Masubuchi and Martin with HY 80,[4] they found cracks always initiated in the HAZ. Although cracks sometimes propagated into the weld metal, there was no evidence in their work of pure weld metal cracks.

Descriptions of the fracture mode alone tell little about the underlying causes of failure. By relating the fracture to the mechanism and structural deformation at elevated temperatures, we are in a better position to suggest remedies applicable to engineering service use.

II

BACKGROUND

Based on the experimental findings of Masubuchi and Martin,[1] a crack in a high strength steel weldment may be characterized as shown in Figure 1. The process of initiation and propagation of a crack is a three stage occurrence:

(1) Initiation of an intergranular microcrack in the heat affected zone.

(2) Growth of the intergranular microcrack to a transgranular macrocrack.

(3) Extension of the macrocrack transgranularly.

To date, most studies regarding weldment cracking has been experimental and empirical. Since a crack is a fracture in metal, it appears that future studies might apply the theories of Fracture Mechanics to weldment cracking. Recently Yurioka[5] studied the third stage of Figure 1 with regards to crack growth and propagation by use of fracture mechanics, but the problem of the total crack propagation has yet to be attacked. Thus the total crack must be analyzed before such studies can begin.

Mechanisms of intergranular hot cracking have been suggested by several investigators.[2] Basically these investigators propose that when a metal is heated to a

temperature just below the solidus temperature, some form of liquation takes place along grain boundaries. During subsequent cooling, tensile stresses are produced by the differential contraction of the metal and result in strain concentrations occurring in the grain boundaries. Intergranular cracks will result when the material can no longer withstand the stresses and/or strains.

Most of the theoretical research on the fracture of metals at high temperature has been concerned with creep fracture. Both creep and stage one of the weld crack are intergranular, but they have different characteristics. The temperature range for creep is generally well below the solidus temperature and grain boundary liquation is not a necessary condition for creep fracture. Additionally, the strain rate for creep fracture is much slower than that of weld metal expansion and contraction strains. The problems of each and of the HAZ as an entity itself are compared below:

	Temperature	Time (Minutes)	Mechanism of Failure
CREEP	0.5 Tm	10^{2+}	grain boundary shearing; void coalescence
WELD METAL	Tm+	10^{0}	Liquation
HEAT AFFECTED ZONE	Tm-	10^{-1}	Incipient melting grain boundary sliding

The metallurgical studies of fracture are fairly well summarized in several reviews of the problems of fracture of metals.[6-12] Only in the past 10-15 years has there been a concentrated effort to combine the various problems associated with weld fracture within one study.

Masubuchi and Martin[4] attacked the problem directly in HY 80 steel and offered a hypothesis explaining cracking in HY 80 weldments based on experimental work, a survey of then current welding problems and metallurgical theory. However their hypothesis also stopped at stage one intergranular cracking. This work will now attempt to correlate all three stages of weldment cracking, using HY 130 steel, and propose a theory for the stage two transition from intergranular to transgranular fracture. It is well to note at this point that there are other mechanisms that cause the transition from intergranular to transgranular fracture, such as crack velocity considerations. However, this work will concentrate on the metallurgical aspect of the transitional problem.

III

OBJECTIVE AND PLAN OF WORK

The objective of this research will be to verify and extend the results of Masubuchi and Martin using HY 130 steel. A review of the mechanisms leading to their proposal for intergranular cracking will be reviewed and correlated to actual crack extension in the transgranular mode. Since their work concluded with the proposal for intergranular cracking only, this work will be directed towards developing the theory for cracking in the transition from intergranular to transgranular cracks.

To do this, a previously welded specimen will be sectioned and examined to identify the type of cracking present. The specimen chosen was a Navy Circular Fillet Welded specimen, a self restrained specimen. The specimen is used to evaluate crack susceptability in the weld metal and HAZ.[6] In this work, the circular fillet welded specimen will be examined metallographically to locate a HAZ crack and then identify it.

In order to determine the various mechanisms of fracture of HY 130 during welding, a test program was established using a GLEEBLE device to simulate the various thermal cycles the HAZ of a weldment would experience in an actual weld cycle. Five test cycles were programmed on the Gleeble device as follows:

Cycle 1. Heating to 2500°F, cooling immediately to the test temperature and fracturing. This cycle attempts to simulate the conditions in the weld HAZ near the fusion line.

Cycle 2. Heating directly to test temperature and then fracturing. This is an attempt to remove the influence of liquation and/or segregation at high temperature.

Cycle 3. Heating to 2500°F and holding for 15 to 20 minutes, then cooling to test temperature and fracturing. This is to determine the effects on the fracture characteristics by holding the specimen at 2500°F for a time longer than ordinarily used.

Cycle 4. Heating to 2650°F, cooling to room temperature, heating to test temperature and fracturing.

Cycle 5. Heating to 2650°F, cooling to room temperature, then heated to 2500°F, cooling to test temperature and fracturing.

Cycles 4 and 5 are used to determine whether liquation or segregation plays an important role in the embrittling of steels at high temperature. All five cycles are displayed graphically in Figure 2.

The tests were conducted on specimens machined with their axes either perpendicular or parallel to the principal rolling direction to determine the effect of orientation on the cracking mechanism.

The results of the actual welded specimen cracking investigation were correlated with the results of the Gleeble test. The causes and mechanisms of cracking will be described on the basis of this correlation.

IV

DESCRIPTION OF APPARATUS AND MATERIALS USED

A. APPARATUS

The apparatus used in this study was the model 501 GLEEBLE described in the literature.[13,14] The device subjects a sample to heating and cooling according to a programmed cycle by means of controlled resistance heating. Temperature in the test area is measured and controlled by means of thermocouple which is percussion welded to the midlength of the specimen. The thermocouple wire used in this investigation was platinum - platinum 10% rhodium, necessitated by the application of temperatures in the range 2500°F - 2650°F.

The uniformly heated area at midlength extends approximately 0.05 inch from each side of the thermocouple. The device is capable of consistently reproducing any programmed thermal cycle. In this investigation, the heating/cooling cycles employed simulated those occurring in the heat affected zone of the base metal during cooling.

The device is also capable of applying an axial tensile load up to 10,000 lb to produce straining rates from 0.2 to 20 inch per second at any point in a given thermal cycle.

A view showing a specimen mounted in the jaws is given by Figure 3.

An eight channel VISACORDER was connected to the Gleeble to provide a continuous record of temperature, load and total strain, all as a function of time.

B. MATERIAL

The material for this investigation was supplied by the Naval Ship Research and Development Center, Annapolis. All material was Navy specified HY 130. The chemical composition is given in Table 1. The mechanical properties of the steel in the "as received" quenched and tempered condition are as follows:

Yield Strength 130-145 ksi

Elongation in 2 inches - 15%

Reduction of area - 50% min. Transverse
20% min. Through Thickness

Impact requirements - 60 ft-lb at 0°F and room temperature.

The material used for the high temperature testing was supplied as 1 inch thick plate. The circular fillet specimen was received in the "as welded" condition using current Navy welding specifications and approved filler wire. A view of the circular fillet specimen "as received" is given by Figure 4.

TABLE 1

BASE PLATE MATERIAL CHEMICAL COMPOSITION

HY 130

C	0.12
Mn	0.60-0.90
P	0.010
S	0.015
Si	0.20-0.35
Ni	4.75-5.25
Cr	0.40-0.70
Mo	0.30-0.65
Ti	0.02 Maximum
V	0.05-0.10
Cu	0.25 Maximum

V

PROCEDURE AND RESULTS

PROCEDURE

Specimens for use on the Gleeble were cut from the 1 inch plate with specimen axes oriented either parallel or perpendicular to the principle rolling direction (Figure 5). It was originally planned to test 5 specimens oriented parallel (coded L) and 4 specimens oriented perpendicular (coded T) for each of the five test cycles programmed. However test cycle 3 had to be eliminated because the feedback sensitivity of the Gleeble machine caused excessive temperature oscillations and caused premature melting of the specimen. Thus 7 parallel specimens and 5 perpendicular specimens were used in each of the remaining test cycles.

Following the input program, the central portion of each specimen was heated to a temperature just below melting and allowed to cool to a predetermined temperature and then fractured dynamically in tension at that temperature. The basic heating cycles used are diagrammed in Figure 6. The tension was applied at a constant headspeed of 590 ipm. The load applied was determined by use of load cell connected to the specimen.

Reduction of area of the specimen was used as a measure of ductility.

The fractured surfaces were then sectioned and examined by conventional metallography and scanning electron microscope.

The circular fillet welded specimen was sectioned to reveal the weld HAZ (Figure 7) and examined until a micro-crack was located. The examination was to determine the mode of cracking exhibited by the highly-restrained, full-scale weldment.

RESULTS

A. GLEEBLE testing - The results of the Gleeble testing are given in tabular form in Appendix A and exhibited graphically in Figure 8. Test temperature at fracture is plotted against reduction of area. Examination of the fractured specimen reveals distinct changes in the mode of cracking that caused failure. The changes have a direct correlation to the shape of the reduction of area curve as follows:

- Below 70% reduction of area, fracture was intergranular (Figures 9a,b,c and 10a,b). [The slight irregularity on the grain boundary is due to platelets of FeO on the surface of the grain].

- From 70% to about 83% reduction of area, there is a transition zone where the cracking mechanism is a combination of intergranular and transgranular fracture (Figure 11). Also in this zone there is the appearance of

relatively large MnS inclusions (Figure 12). The MnS and the FeO were identified by an energy dispersal X-ray analyzer on the scanning electron microscope.

· Above 85% reduction of area the fracture mechanism was transgranular. (Figures 13a,b,c).

Grain distortion was observed in all the specimens as evidenced by the slip lines. The slip lines and grain distortion increased with increasing reduction of area.

The orientation of the specimen axis to plate rolling direction (parallel or transverse) had little effect on the high temperature fracture properties. Similarly, there was little difference in properties due to fracturing on heating as compared with the cooling cycle.

B. Welded Specimen examination - The examination of the circular fillet welded specimen revealed several macrocracks at the fusion line in addition to the microcracks which developed in the HAZ adjacent to the fusion line. In every instance investigated there was a MnS inclusion at the center of the macrocracks. The crack culminated with a transgranular "tail" extending outward from the center (Figures 14,15).

VI

DISCUSSION

A general requirement for cracking to occur is that a strain must be applied to the material. However, in the case of weld cracking, the strains in question are highly localized. Typical sources of strain that can be directly associated with welding are:

a) Differential expansion and contraction arising from the application of an intensive local heat source in the form of the welding arc. This means that the solid metal already in the joint vicinity experiences a cycle in which it is rapidly heated and then rapidly cooled all the while it is undergoing plastic strain.

b) In a structural component, plastic strain can occur at a point due to general deformation of the total component. This deformation may be produced either as a direct result of welding or from some other source.

However, the total strain in weldments is generally less than about 5 per cent. Furthermore, the residual stresses induced by welding operations are localized to the weld area and cannot be much higher than the material yield stress. Thus there must be other mechanisms operating that cause cracking in welds.

Other causes of microcracks in welds have been attributed to:

1. low ductility over some intermediate temperature range

2. the presence of non metallic films at grain boundaries

3. the existence of localized melting at grain boundaries during deposition of weld metal during multipass welding.

In some cases, microcracking can result from more than one of the above mechanisms acting simultaneously. In all the cases, the necessary strain to induce cracking is caused by the differential expansion and contraction of the deposited weld metal.

The fracture of metals is greatly more complex than that of completely brittle solids because of an inherent capacity for plastic flow in the metallic bond which gives rise to a number of different fracture mechanisms.

In the discussion of fracture mechanisms, it becomes convenient to build on the fracture mechanisms of single crystals and then formulate fracture of polycrystalline aggregates. As is the case in most practical studies, seldom, if ever, will fracture occur as the result of a

single mechanism. Usually there are two or more mechanisms which are present and can be considered the cause of failure.

Mechanisms of Fracture of Single Crystals [7]

(1) Shear Fracture

Single crystals of certain metals will deform by slip, or the sliding of one part of the crystal over another. Shear fracture occurs when the sliding reaches the point where the two parts of the crystal are completely separated.

(2) Glide, or Glide Plus Twinning

Examples of this mechanism are evident in single crystals of zinc and cadmium. Here the lattice reorientation caused by twinning may place new slip systems in a favorable position relative to the stress axis so that plastic deformation occurs more readily inside the twin. This generally produces necking of the specimen, and the specimen draws to a wedge and separates along a line.

(3) Cleavage

At low temperature, single crystals may fracture by cleavage, or separation perpendicular to some crystallographic plane. No face - centered cubic metal is known to fail by this mechanism.

Fracture of Polycrystalline Metals

In polycrystalline materials, the mode of fracture is inherently linked with the mechanism of flow.[12] Some of the modes of flow of crystalline materials which follow from single crystal considerations are:

Slip which is the motion along parallel planes within a crystal lattice. In general, the "easy glide" plane is that lattice plane which has the highest atomic density. Slip initiates in the crystal whose easy glide plane coincides with the direction of maximum shear.

Twinning is a form of plastic deformation which, like slip, occurs as the result of applied stress. Here parallel planes move relative to each other in parallel directions so that the lattice on one side of the twinning plane is a mirror image of that on the other side.

In polycrystalline materials, the crystals or grains usually have a random orientation and are separated by boundaries. Thus the boundaries and adjacent grains act to restrict the deformation of the grain. Small deformation of the grain can be produced by slip along five independent slip systems. In lattices with less than this number of slip systems, or where there is a restriction to grain boundary motion, a general deformation cannot be produced without rotation, elastic distortion and bending of the slip plane.[12]

Boundary flow or deformation of the grain boundary. Here the grain boundary material exhibits viscous like behavior as long as the strains are small so that the grains are not caused to rotate or move enough to touch the adjoining grain. As the temperature is increased, the grain boundary material deforms even more easily than that of the grains and a greater share of flow takes place in the grain boundaries.

Substructure formation is a mechanism by which the grains break up into a subgrain structure. The orientations of the subgrains are substantially different from that of the parent grain. The boundaries of these subgrains behave like grain boundaries.

Essentially, slip and grain boundary sliding are the chief methods by which a polycrystalline specimen achieves elongation during deformation at high temperatures. Ductility is that property of a metal which enables it to permanently alter its shape under the action of an applied stress.

Ductile, transgranular failures appear in most metals and alloys at temperatures below the equicohesive temperature. Although originally defined as a material property of a strength balance between cohesion at the grains and within the grain, equicohesive temperature is used to describe the temperature at which the type of fracture for a

given metal changes from predominantly transgranular to predominantly intergranular; the higher temperature regime corresponding to intergranular failure, provided of course that strain rates are suitable.

There have been various explanations[16-21] of the initiating factors surrounding intergranular fracture in metals based on dislocation theory. Zener[16,17] used the term "micro-mechanism of fracture" and noted that grain boundaries behave in a viscous manner and as such can relax stresses imposed upon them. The junction point of three grains can not provide relative movement and these triple points can experience high degrees of stress concentration. An example of a Zener crack is given by Figure 16. The tensile stress is taken as applied along an axis normal to the grain boundary, BB'. This boundary is intersected by another grain boundary AA', which has a shear stress component of the tensile stress. Any relaxation of the shear stress across the boundary AA' results in an extremely high tensile stress in the corner AA'B'. When this stress reaches an intrensic fracture strength, a microcrack will form. Further propagation of the crack may or may not occur depending on the initial crack opening, further boundary sliding, magnitude of stress applied, and further slip.

Cottrell[18] and Petch[17] have further formalized the dislocation theory of the initiation of fracture of metals based on the gliding together of dislocations lying on intersecting planes. The principle mechanisms are shown in Figure 17.

Chang and Grant[21] have given further examples of initiation of intergranular cracks with some interesting additions (Figure 18). Types (a) and (b) show intergranular cracks resulting from sliding on adjacent boundaries with cracking occurring on the non-sliding boundary. Type (c) shows an intergranular crack forming on the sliding boundary itself. Further, more complex configurations were noted, but are merely combinations of the three types given.

From Grant's observations, it is noted that both stationary and sliding boundaries can crack as a result of the interactions.

From the above mentioned theories and using the experimental evidence found in this work, it appears that the initiation of intergranular cracking is indeed as depicted by Masubuchi and Martin and follows directly from the works of Zener and Chang and Grant. As shown by Figure 10 and characterized by Figure 19, the stress concentration at the triple point caused by sliding results in intergranular fracture under small general deformation. The sliding along

the grain boundaries may take place at low stresses since the high temperature (> 2300°F) causes the grain boundaries to behave in a viscous manner with a very low flow stress.

Inasmuch as the formation of microcracks is restricted to a relatively small portion of the grains despite the marked increase in stress in the vicinity of the microcrack, the local stress cannot be the only factor playing an important part in the process of transgranular crack propagation. It is suggested that transgranular cracking occurs when the appearance of precipitation particles stiffen the grain boundaries to oppose further grain sliding and prevent stress relaxation. From the results of the Gleeble tests and the welded specimen examination, relatively large (10μ inches in a 60μ inch grain) inclusions of MnS were found at the center of the microcracks which grew to macrocracks. These inclusions act as a soft "plastic hinge" absorbing the local plastic strain and preventing the grain boundary from further sliding. The stress concentrations that remain after the initial intergranular crack opening appear to be controlled by the MnS inclusion and further cracking takes place transgranularly from this plastic hinge (Figure 20).

The precise stage of the precipitation process at which transgranular cracking occurs is difficult to determine, but there is no evidence that transgranular cracking takes place

prior to that at which the precipitates can surface to the grain boundary.

VII

SUMMARY AND CONCLUSIONS

- The Gleeble was able to reproduce the conditions exhibited by actual weldments during welding and the subsequent cracking of the weldment due to the welding process.
- Cracking in HY 130 weldments was confirmed to be a three stage phenomena. The first stage takes place in the HAZ at or near the fusion line where the temperature is just below the melting temperature. Cracking was characterized by intergranular separation with little or no grain distortion. These cracks extended for one to two grain diameters (approximately 150μ inches) and were evenly distributed throughout the HAZ.

The second stage of cracking is the intersection of the intergranular cracking with a relatively large sulfide inclusion. This inclusion locks the grain boundary from further sliding and causes a stress concentration at the grain boundary.

The third stage is the release of the grain boundary stress concentration by way of a transgranular crack opening.

- The MnS inclusions found appear to be large agglomerations that readily concentrate the stresses transmitted by the grain boundaries. These agglomerations absorb the local strains and act as a plastic hinge to cause further cracking to take place transgranularly.

The ultimate failure of specimens require the nucleation, growth, and intersection of many cracks. There are numerous factors controlling these conditions, but the most noticeable are: the length of the initial intergranular cracks and their density, the degree of plastic deformation of the grains, and the strain induced precipitates acting on the boundaries.

VIII

FUTURE WORK

This investigation has shown that the presence of rather large agglomerations of MnS act as a nucleus from which transgranular cracking takes place. To reduce the incidence of transgranular cracking the following steps are recommended:

1. Reduce the sulfur content even further than the 0.015 per cent presently allowed. The actual amount of reduction would have to be experimentally determined.

2. Since the MnS appears as such a large precipitate, steps must be taken in the ingot processing procedure to break up the sulfide concentration and disperse the sulfides to form a more homogeneous state within the final plate.

Furthermore, research sould be done within the processing procedure to determine the circumstances surrounding the formation, distribution and function of the MnS inclusions.

Finally, this work recognized MnS precipitates as a means by which the cracking mechanism can transform from the intergranular mode to the transgranular mode. Additional research is necessary to determine what other precipitate forms can react in the same manner.

BIBLIOGRAPHY

1. Masubuchi, K. and Martin, D. C., "An Analytical Study of Cracking in Weldments," a report prepared for Battelle Memorial Institute.

2. Kammer, P.A., Masubuchi, K., and Monroe, R. E., "Cracking in High Strength Steel Weldments - A Critical Review," DMIC Report 197, February 7, 1964.

3. Berry, J. T., and Allan, R. C., "A Study of Cracking in Low-Alloy Steel Welded Joints," WELDING JOURNAL, Research Supplement, 105s - 116s, March 1960.

4. Masubuchi, K., and Martin, D. C., "Mechanisms of Cracking in HY-80 Steel Weldments," WELDING JOURNAL, Research Supplement, August, 1962.

5. Yurioka, N., "Analysis of Permissable Weld-Flaw Sizes in High-Strength Steel Pressure Hulls for Deep Submersibles," October 29, 1971, M.I.T. Department of Ocean Engineering (unpublished report).

6. Vagi, J. J., Meister, R. P. and Randall, M. D., "Weldment Evaluation Methods," DMIC Report 244, August 1968, p35.

7. Spretnak, J. W., "A Summary of the Theory of Fracture in Metals," DMIC Report 157, August 7, 1961.

8. Petch, N. M., "The Fracture of Metals," Progress in Metal Physics, Vol. 5, Pergamon Press Ltd., p1.

9. Fracture (Swampscott Conference), edited by B. L. Averback, D. K. Felbeck, G. T. Hahn, and D. A. Thomas, The Technology Press, M.I.T., 1959.

10. Mechanical Behavior of Materials at Elevated Temperature, edited by J. E. Dorn, McGraw Hill Book Co., Inc., N.Y. p.218-269, 1961.

11. Tetelman, A. S., and McEvily, A. J., Jr., "Fracture of High Strength Materials," FRACTURE, edited by H. L. Liebowitz, Academic Press, N.Y., 1969.

12. Manjoine, M. J., "Fracture at Elevated Temperatures: Mechanical Aspects,: Fracture of Engineering Materials, ASM, Metals Park, Ohio, 1964.

13. Savage, W. F., "Apparatus for Studying the Effects of Rapid Thermal Cycles and High Strain Rates on Elevated Temperature Behavior of Materials," JOURNAL OF APPLIED POLYMER SCIENCE, VI, 21, p303-315.

14. Nippes, E. F., Savage, W. F., and Grotke, G., "Further Studies of the Hot Ductility of High Temperature Alloys," Welding Research Council Bulletin No. 33, February 1957, p32.

15. Baker, R. G., and Newman, R. P., "Cracking in Welds," METAL CONSTRUCTION & BRITISH WELDING JOURNAL, February 1969, p1s-4s.

16. Zener, C., "The Micro-Mechanism of Fracture," Fracturing of Metals, ASM, Cleveland (1948) p3.

17. Zener, C., Elasticity and Anelasticity of Metals, University of Chicago Press, Chicago, 1948, p158.

18. Cottrell, A. H., "Theoretical Aspects of Fracture," Fracture, edited by B. L. Averbach, D. K. Felbeck, G. T. Hahn, and D. A. Thomas, The Technology Press, M.I.T., 1959, p20.

19. Gifkins, R. C., ACTA METALLURGICA, Vol. 4, January 1956, pp98-99.

20. Chen, C. W., and Machlin, E. S., "On the Mechanism of Intercrystalline Cracking," ACTA METALLURGICA, Vol. 4, November 1956, pp655-656.

21. Chang, H. C., and Grant, N. J., "Mechanism of Metercrystalline Fracture," TRANSACTIONS AIME, Vol. 26, May 1956, pp544-551.

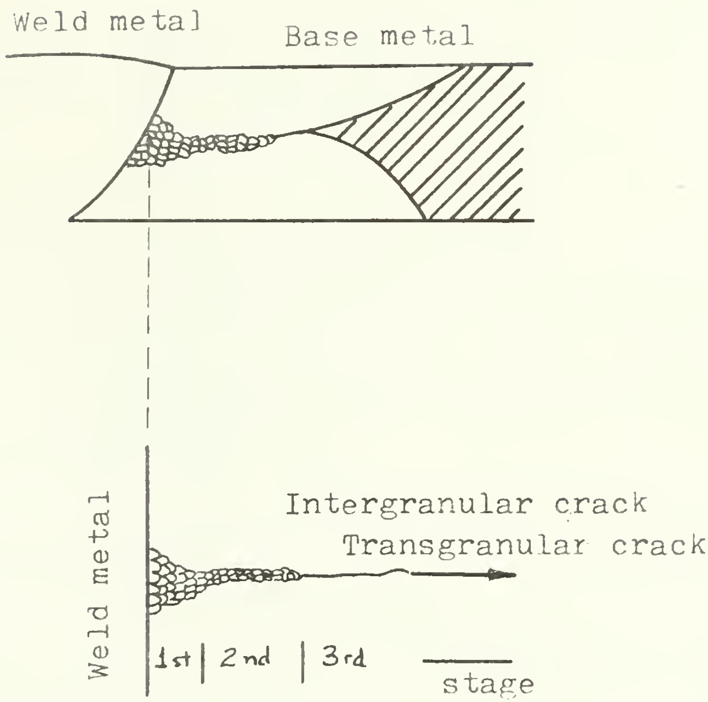


Figure 1. Schematic figure of cracking in a weldment. The crack may occur in various directions. (Masubuchi and Martin)

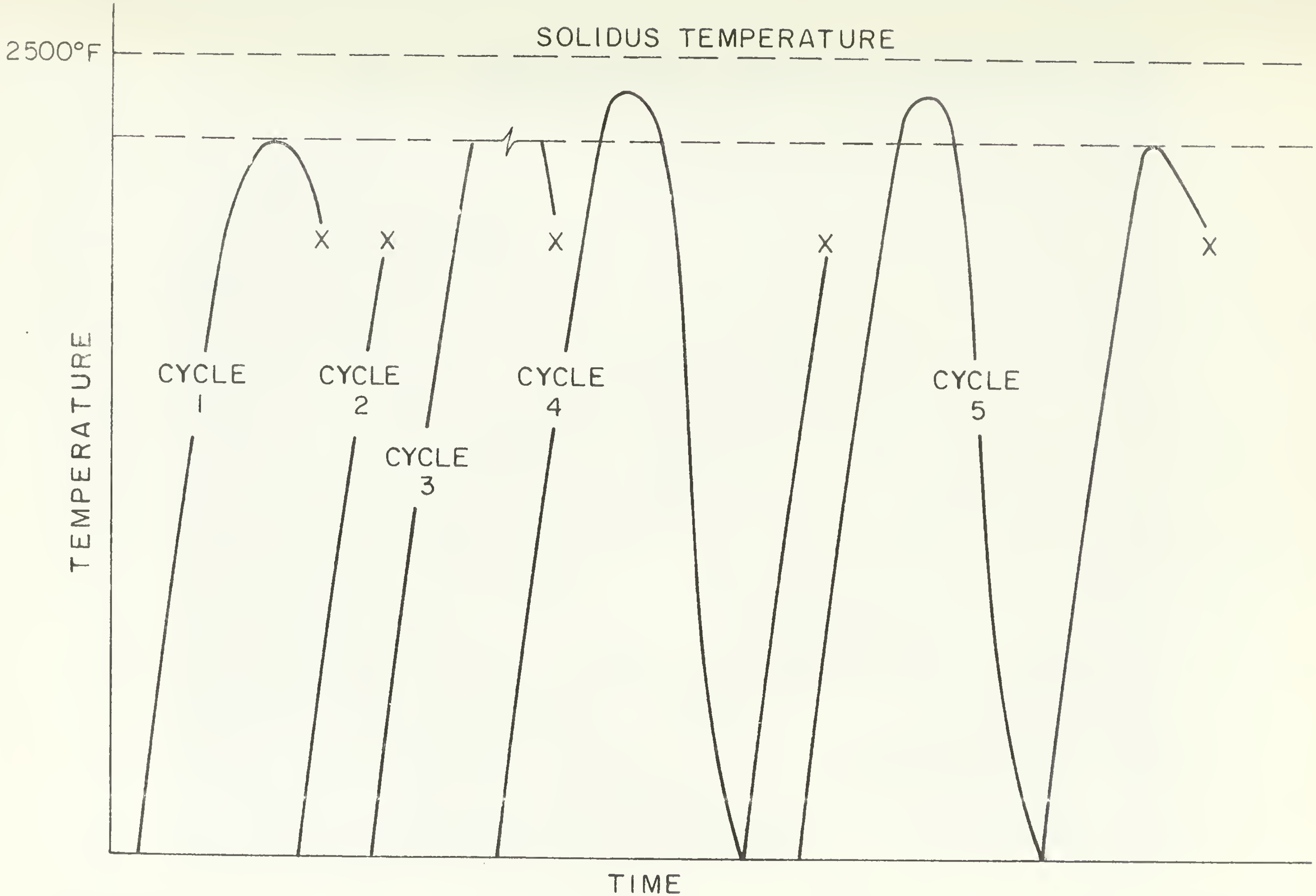


FIGURE 2 THERMAL CYCLES USED IN FRACTURE TESTS WITH THE GLEEBLE DEVICE

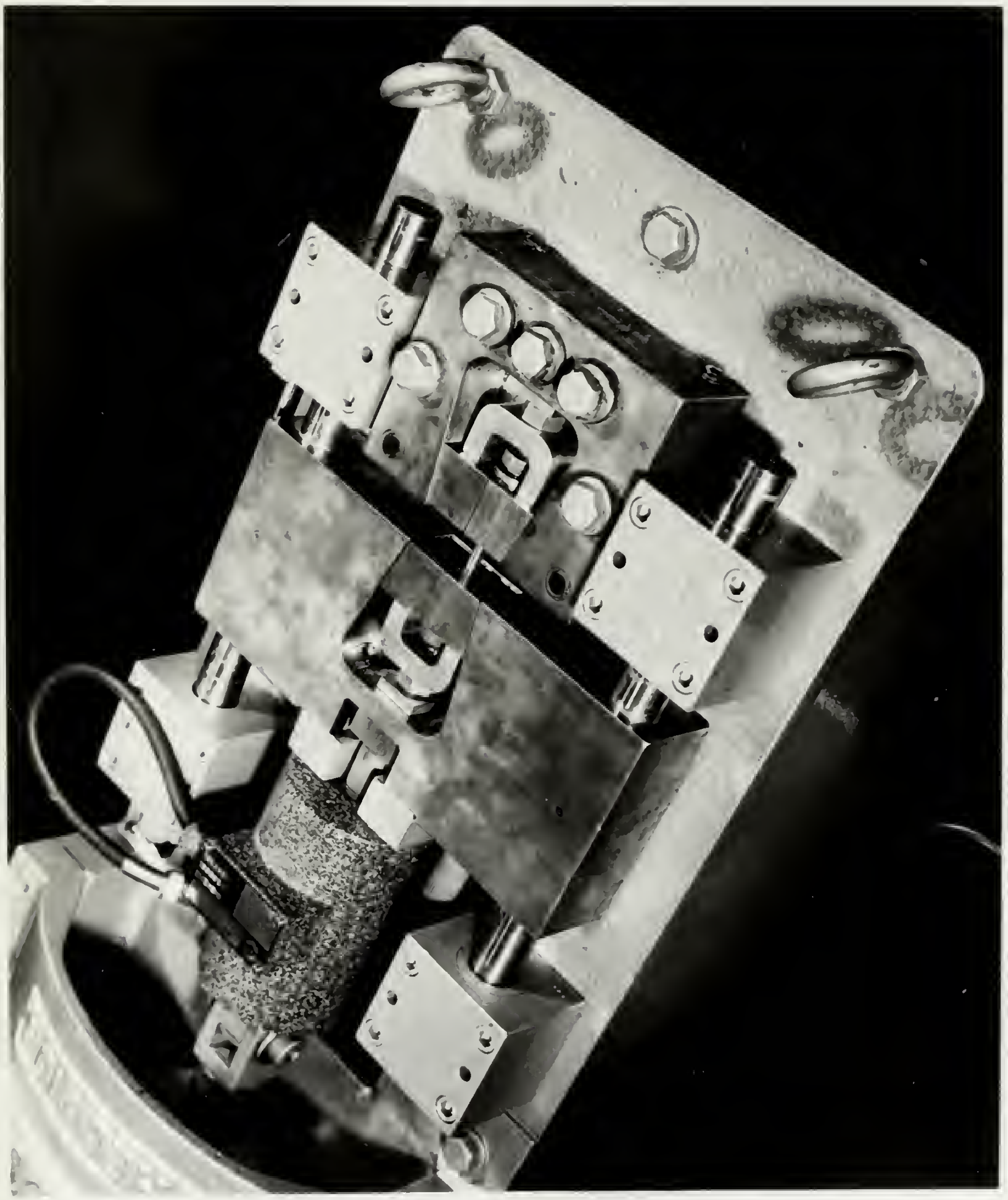


Figure 3. A View of a Specimen Mounted in the Jaws of the Gleeble

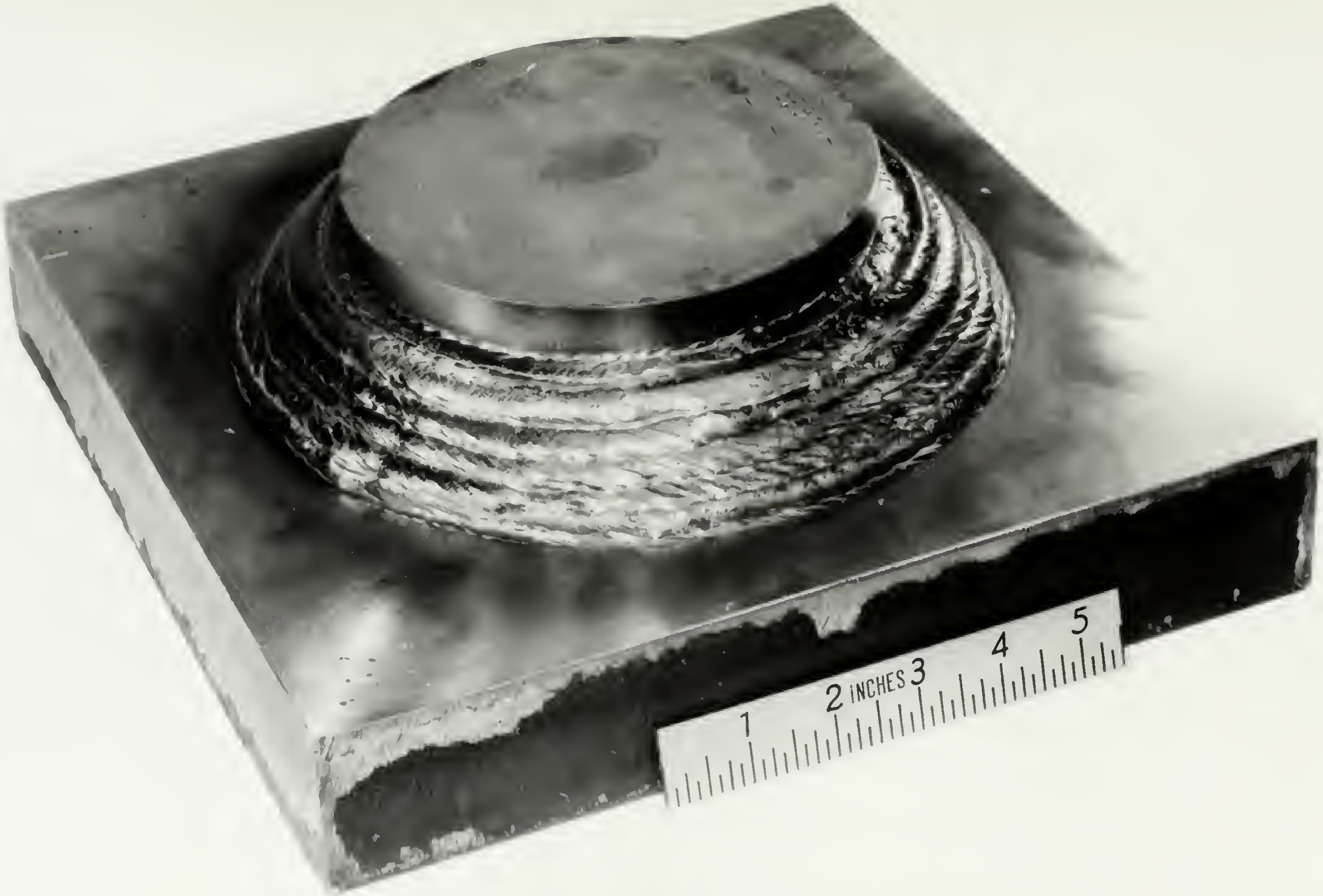


Figure 4.-Naval Circular Fillet Welded Specimen.

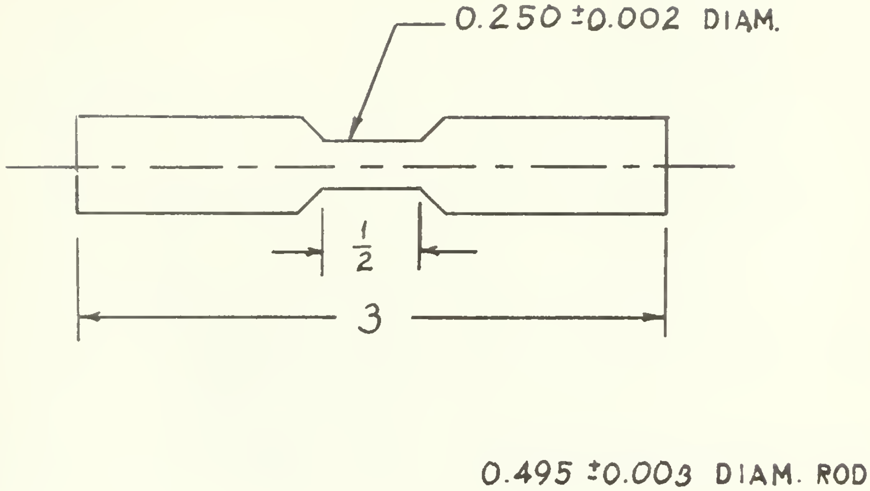


Figure 5. Specimen used for Gleeble tests.

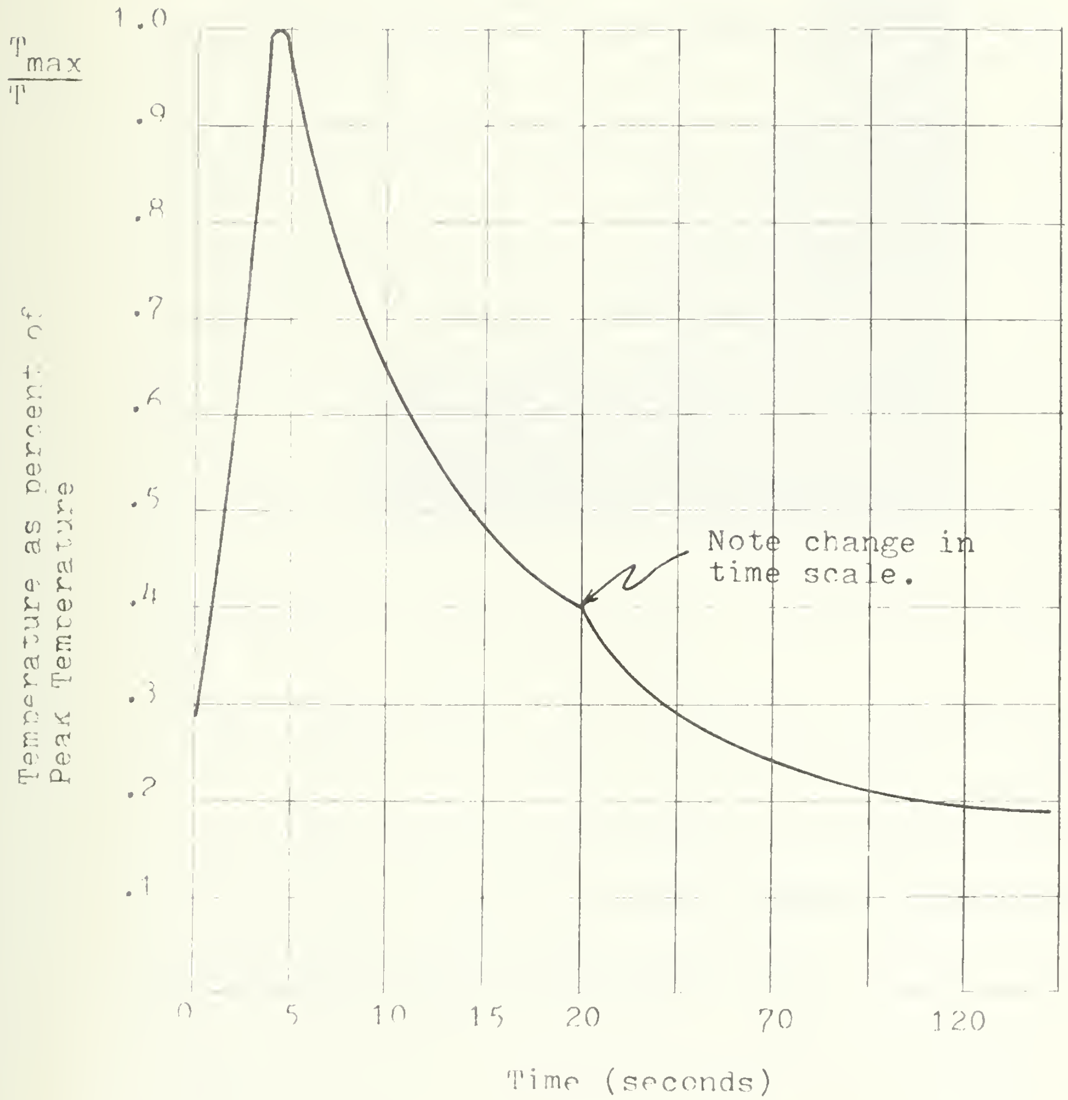


Figure 6. Basic heating/cooling cycle programmed on the Gleeble device.

Figure 7

A view of the Heat Affected Zone of the Circular Fillet Welded Specimen.

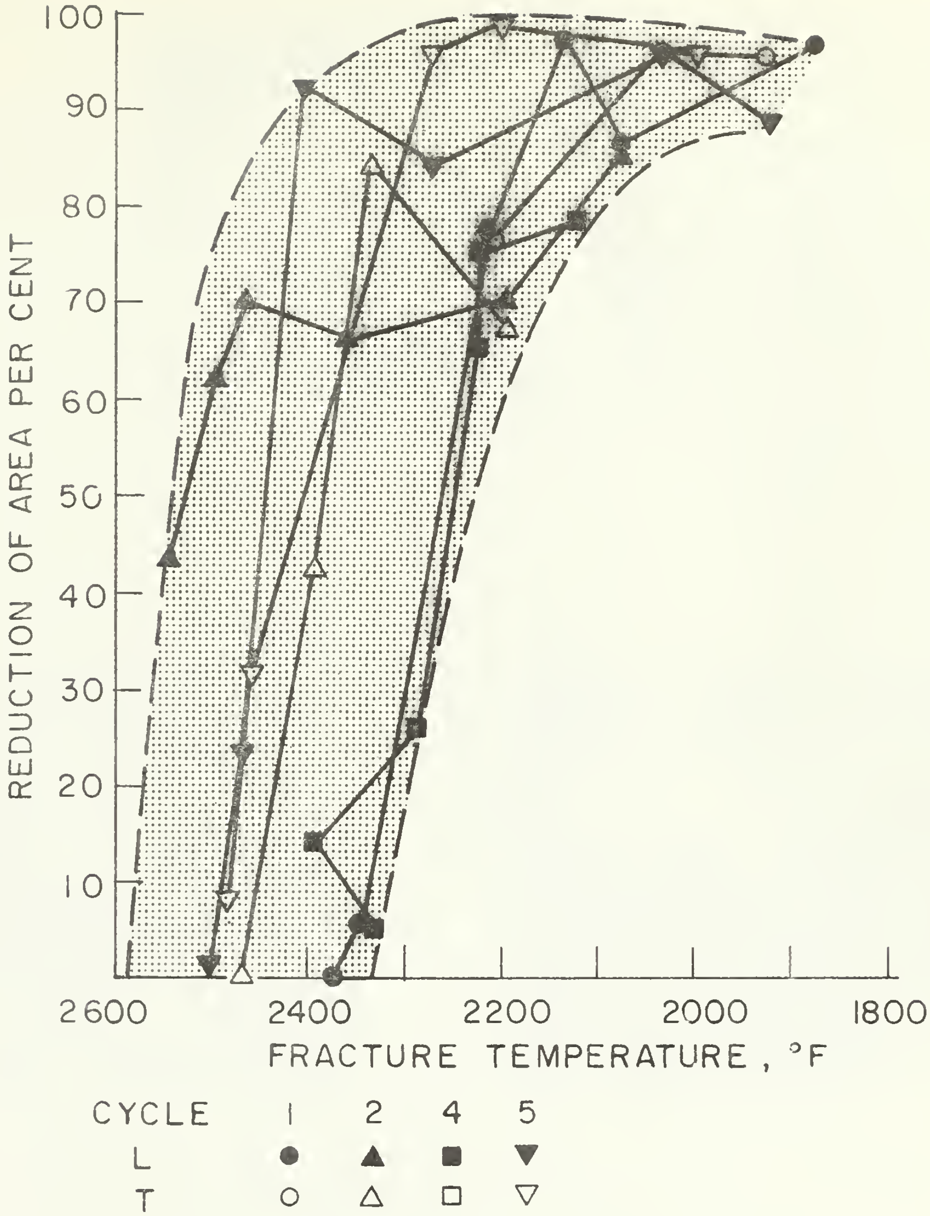
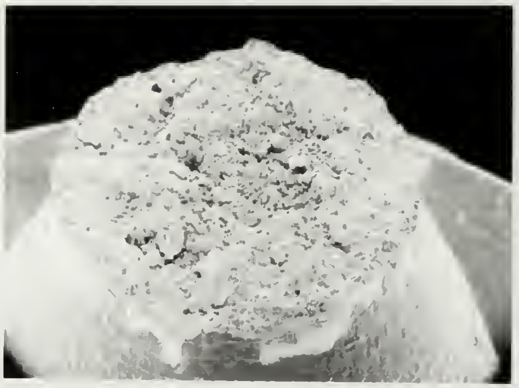


FIGURE 8 RESULTS OF THE GLEEBLE TESTS

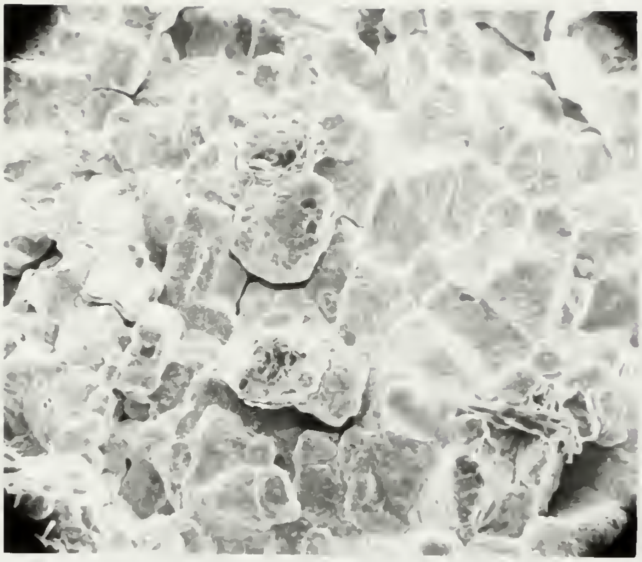


(a) 23X

(b) 1160X

(c) 2375X

Figure 9. Microhpotograph from the Scanning Electron Microscope (SEM) of specimen 2L6 showing intergranular cracking.



(a) 120X

(b) 240X

Figure 10. Microphotograph from the Scanning Electron Microscope (SEM) of specimen 5L6 showing intergranular cracking.

Figure 11. Microphotograph from the SEM viewing a mixed fracture of intergranular and transgranular cracking. (2250X)

Figure 12. Microphotograph from the SEM observing the MnS inclusion found at the grain boundaries. (2375X)

(a) 22X

(b) 2330X

(c) 5700X

Figure 13. Microphotograph from the SEM viewing transgranular cracking of specimen 1L7.

Figure 14. Transgranular Crack in NCFW Specimen. (500X)

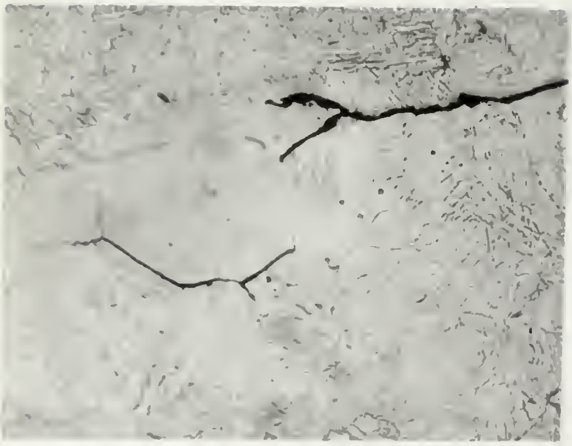


Figure 15. Intergranular and Transgranular Cracks in NCFW Specimen. (1000X)

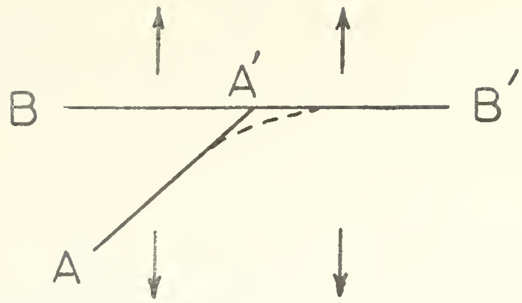
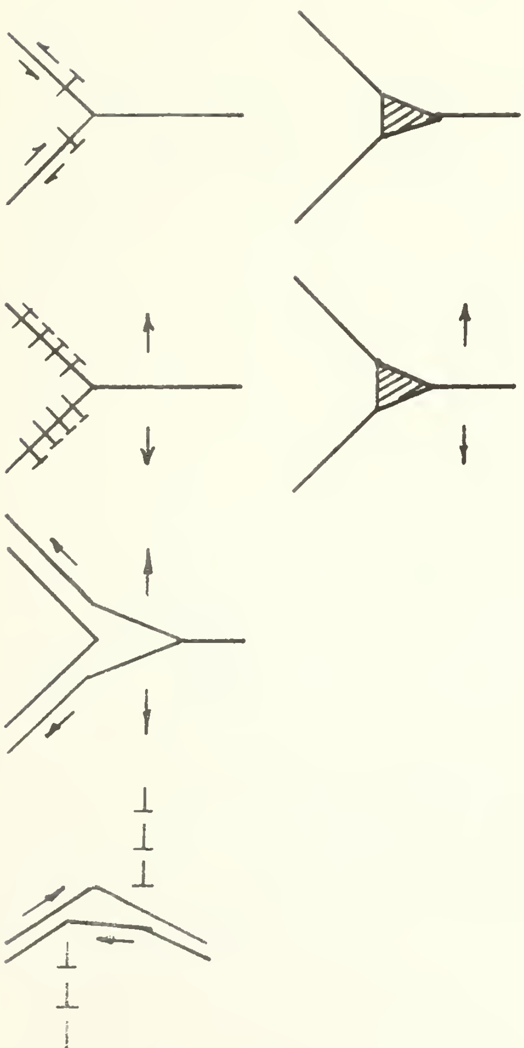


Figure 16. Crack formed by shaer stress relaxation along the grain boundary. (Zener)



a) Coalescence of two slip dislocations to form a crack on a cleavage plane.

b) Coalescence of two slip bands to form a crack.

c) Crack resulting from shear on two bands.

d) Crack formed at tilt boundary.

Figure 17 (Cottrell)

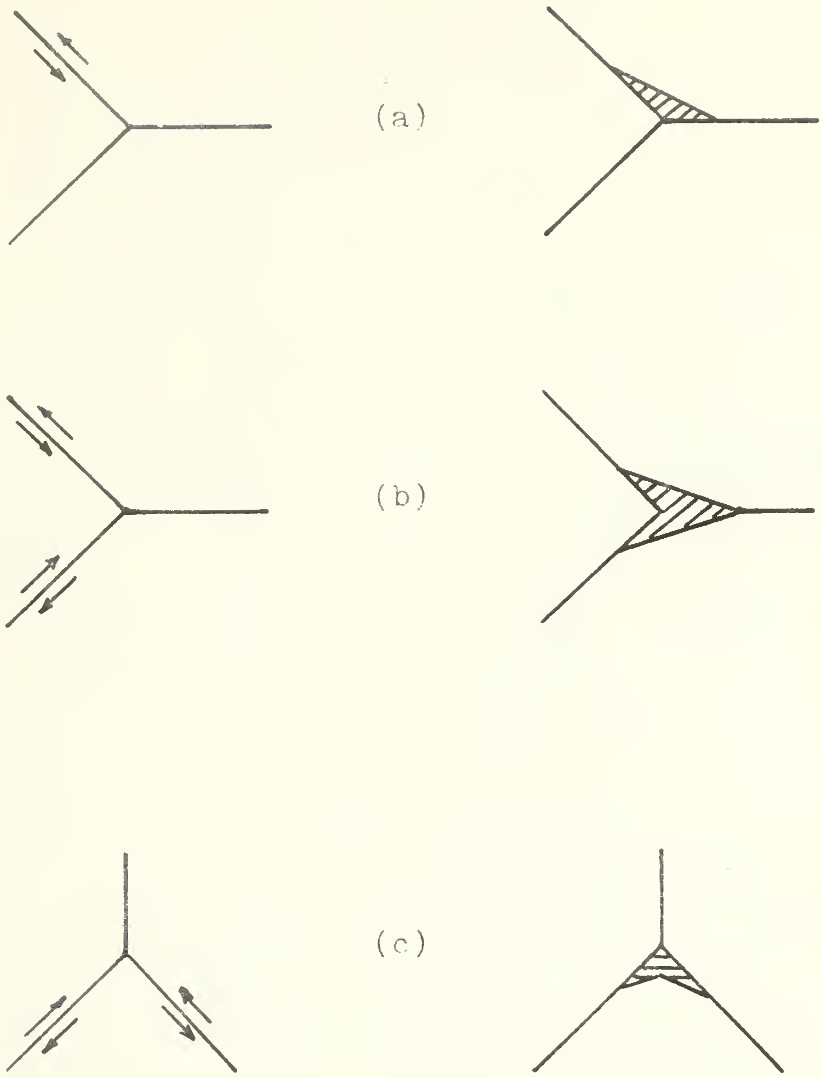


Figure 18. Schematic representation of wedge crack formation by grain boundary sliding. (Chang and Grant)

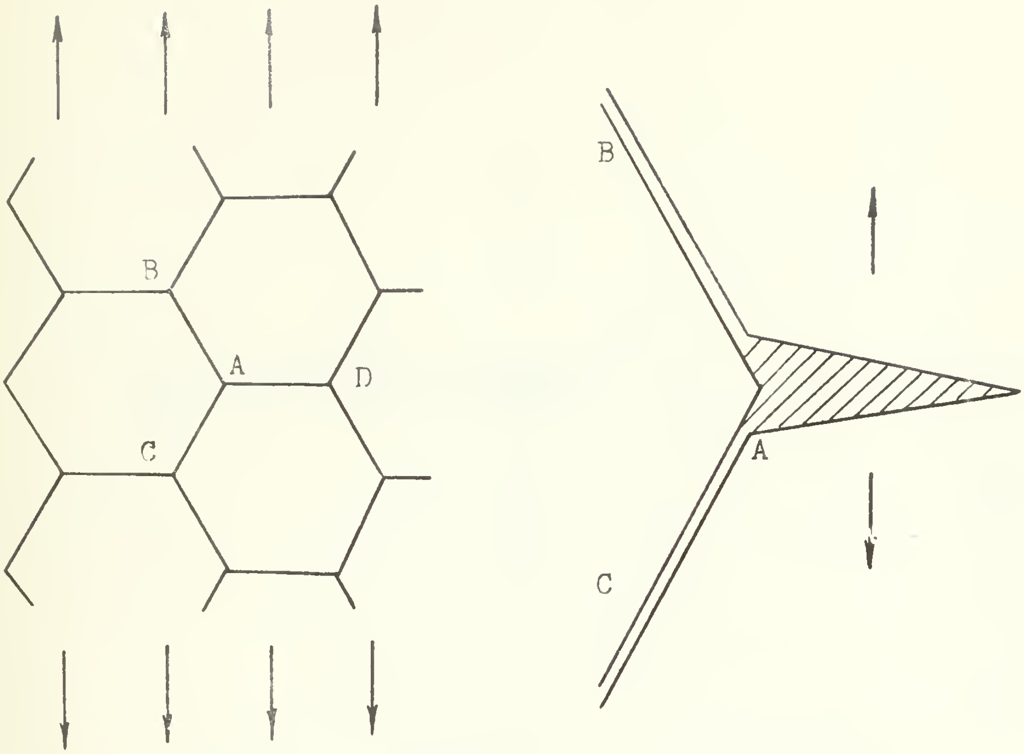


Figure 19 Formation of intergranular cracking A (left)- idealization of stress condition; B (right)- intergranular cracking resulting from slipping along two intersecting grain boundaries.(Masubuchi and Martin)

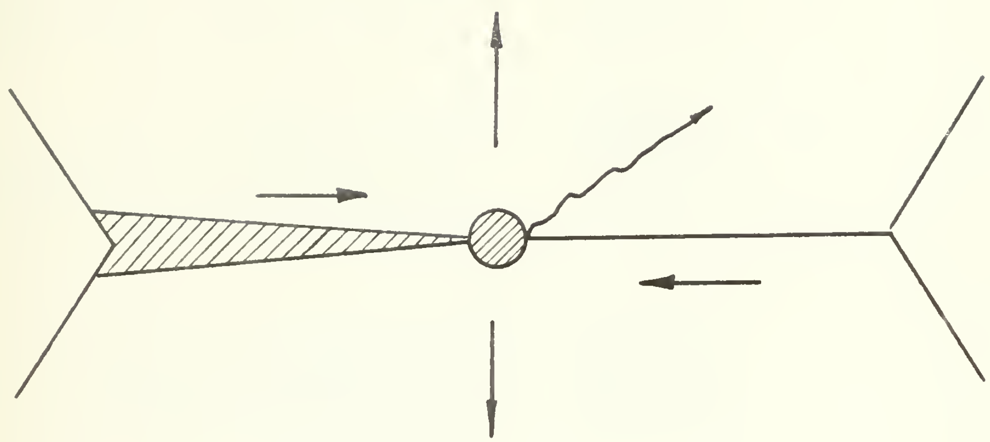


Figure 20 Formation of a Transgranular crack resulting from the intersection of an Intergranular crack with an inclusion.

APPENDIX

RESULTS OF GLEEBLE TESTS

SPECIMEN	TEMP PEAK	TEMP FRACTURE	AREA INITIAL	AREA FINAL	$\frac{A_o - A_f}{A_o}$ R.A.
1L1	2496	2376	.049087	-----	0
1L2	2496	2396	.049087	-----	0
1L3	2495	2346	.049876	.04714	5.478
1L4	2491	2210	.048695	.010935	77.54
1L5	2476	2135	.049087	.00125	97.44
1L6	2491	1874	.04948	.00159	96.78
1L7	2521	2074	.049876	.006903	86.16
1T1	2491	2210	.048305	.011499	76.19
1T2	2536	2035	.048695	.002376	95.12
1T3	2521	1927	.049087	.00246	94.98
1T4	MELT				
1T5	MELT				

CYCLE 1

	TEMP		AREA		$\frac{A_o - A_f}{A_o}$
SPECIMEN	PEAK	FRACTURE	INITIAL	FINAL	R.A.
2L1	2542	2542	.04948	.02776	43.8
2L2	2497	2497	.049087	.01863	62.05
2L3	2361	2361	.049876	.0165	66.9
2L4	2466	2466	.049876	.0145	70.87
2L5	2295	2295	.049087	-----	-----
2L6	2195	2195	.049876	.0147	70.44
2L7	2074	2074	.048695	.00693	85.75
2T1	2466	2466	.040115	.040115	0
2T2	2391	2391	.04948	.02835	42.69
2T3	MELT				
2T4	2195	2195	.048305	.0156	67.68
2T5	2331	2331	.049087	.00785	93.99

CYCLE 2

	TEMP		AREA		$\frac{A_o - A_f}{A_o}$
SPECIMEN	PEAK	FRACTURE	INITIAL	FINAL	R.A.
4L1	2648	2256	.049087	-----	----
4L2	2642	2391	.04948	.042638	13.82
4L3	2642	2331	.049087	.04637	5.52
4L4	2642	2286	.049087	.036305	26.04
4L5	2642	2226	.04948	.012867	73.9
4L6	2648	2226	.049087	.01697	65.4
4L7	2642	2120	.049876	.01056	78.8
4T1	MELT				
4T2	MELT				
4T3	2642	2497	.049087	-----	----
4T4	2642	2195	.04948	.0114	76.76
4T5	2642	2572	.047916	.047916	0

CYCLE 4

	TEMP		AREA		$\frac{A_o - A_f}{A_o}$
SPECIMEN	PEAK	FRACTURE	INITIAL	FINAL	R.A.
5L1	2645	1921	.049087	.005808	38.16
5L2	2645	2035	.049087	.002123	95.67
5L3	2655	2271	.049087	.007697	84.3
5L4	2652	2406	.048695	.003631	92.5
5L5	2655	2466	.049087	.03766	23.26
5L6	2650	2500	.04948	.048695	1.58
5T1	2627	1998	.04948	.00237	95.19
5T2	2642	2200	.04948	.000907	98.165
5T3	2642	2271	.048695	.002206	95.46
5T4	2625	2451	.048305	.033006	31.67
5T5	2642	2482	.048305	.044488	7.90

CYCLE 5

Thesis
B3332 Becker
145673
Cracking mechanisms in HY-130 steel weldments.